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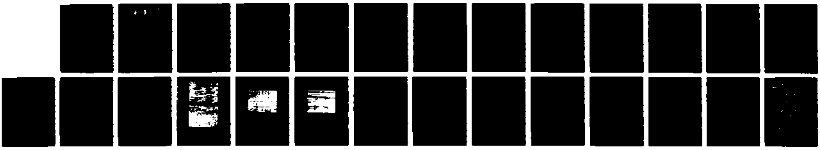

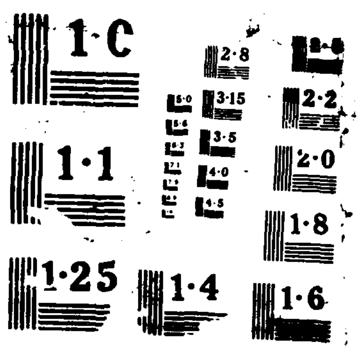
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This report describes progress made in extensive deformation processing of two phase solids, the strengths developed in such materials and their elevated temperature stability. Various combinations of metals can be cold worked to yield high strength, high toughness in situ composites. The strengths they manifest depend on composite constituents, microstructural scale and the extent of cold work applied to them. Some alloys are reasonably resistant to thermal softening. However at very high temperatures they lose their composite like morphology as a result of surface energy effects.

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PROCESSING, STRUCTURE AND PROPERTIES OF HEAVILY DEFORMED IN SITU COMPOSITES

FINAL REPORT

T. H. Courtney and J. K. Lee

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Nickel-tungsten

Silver-Nickel Copper-Iron Silver-Copper Silver-Iron Copper Iron

I. Introduction

Copper-Niobium

This report describes the results of work conducted on the properties, structure, processing and thermal stability of heavily cold worked two phase metallic alloys. These are also known as heavily deformed in situ composites (HDISC). The primary reason for studying these kinds of materials is that two phase materials work harden more rapidly than do single phase ones (1-6). In fact, when subjected to true deformation strains greater than about three ~~(≈95% R.A.)~~, strengths of two phase combinations usually significantly exceed those expected by volumetrically averaging the strengths developed in the single phase constituents subjected to the same strains. At true deformation strains on the order of ten ~~(≈99.996% R.A.)~~, strengths approaching the theoretical ones are obtained in some two phase metal combinations ~~(3,6)~~. This enhanced strain hardening has been observed in a number of two phase metal alloys. Examples include Ag-Cu ~~(3)~~, Ag-Fe ~~(2)~~, Cu-Fe ~~(5)~~, Cu-Nb ~~(4,6)~~, Ag-Ni ~~(5)~~, Cu-Cr (5,7) and Ni-W (8,9).

There is another reason for studying these materials. In particular, their fabrication takes place by relatively straightforward means, and thus they offer the potential for developing relatively inexpensive metal matrix composites. Following primary consolidation (e.g. casting, solid state sintering and liquid phase sintering have all been used), the material is subjected to extensive cold deformation by one or a combination of deformation processes (e.g. swaging, rolling or drawing). This fabrication produces a microstructure which can be called composite like; that is , deformation changes the shapes of the phases consistent with the macroscopic shape change. For example, rolling produces plate shaped dispersed phases (9), whereas drawing typically produces a fibrous dispersion (3-7).

As noted, this program has focused on several areas of research relative to HDISC; processing, properties and their high temperature stability. The results of this program are discussed in separate, following sections on each of these topics. In addition, ancillary research relative to the general area of two phase solids has also been conducted and this is reported on in the last section of this report.

II. Properties

In this program we have concerned ourselves with the reason(s) two phase solids work harden as they do. We have developed a model for the high strain work hardening behavior of such alloys (5,10). The model is a phenomenological one, but is nonetheless based on physically plausible principles. In particular, we believe there are inherent strain incompatibilities across interphase boundaries in a two phase alloy that are greater than those existing across grain boundaries in a single phase polycrystal. The reason for this is

that strain incompatibilities across grain boundaries in a single phase material are only crystallographic in nature, but they arise from both crystallographic effects and the inherently different flow behavior of the phases in two phase solids. Geometrically necessary dislocations can provide the grain boundary accommodative flow in single phase materials, and they can do likewise for interphase boundary incompatibilities in two phase ones (11). For reasons just stipulated their number should be greater in two phase solids.

Our model assumes that geometrical dislocations, once introduced into the material, behave in the same way with respect to multiplication and recovery processes as do ordinary or "statistical" dislocations which are generated during cold working of single phase solids. On this basis we can write an expression (5,10) for the change in dislocation density (ρ) with strain (ϵ) in, say, phase A of the two phase solid as

$$\frac{d\rho_A}{d\epsilon} = C_1\sqrt{\rho_A} - C_2\rho_A + \frac{P_A K}{V_A D} \qquad (1)$$

In Eq. (1), ρ_A represents the total dislocation density in phase A irrespective of whether the dislocations are geometrical or statistical in origin. The first two terms on the right hand side of Eq. (1) represent multiplication and annihilation/recovery processes, respectively (12,13); the latter are responsible for the often observed "close to" saturation hardening found in single phase solids at large deformation strains. The last term on the right hand side of this equation represents the generation of geometrical dislocations arising from strain accommodation across interphase boundaries. The term K in it is a measure of the inherent strain incompatibility between the phases (e.g. K=0 for a single phase material). The term P_A represents the fractional partitioning of geometrical dislocations between the phases; i.e. $P_A + P_B = 1$ where it is expected that if A is the "softer" phase, dislocations will partition preferentially to it so that $P_A > 0.5$. The term D in the denominator of this term is the interphase spacing so that V_A D is the distance over which the inherently different displacements of the phases must be accommodated by virtue of the geometrical dislocations. It has been found that the interphase spacing varies with deformation strain (13,14) as

$$D = D_o \exp[-\epsilon/2] \qquad (2)$$

where D_o is the interphase spacing prior to deformation processing. The relationship, Eq. (2), has been found to hold for both axisymmetric (e.g., wire drawing) and plane strain (e.g., rolling) macroscopic deformation processes.

The flow stress of each of the phases in the composite is assumed represented by an equation of the form

$$\sigma = \sigma_o + \alpha M G b \sqrt{\zeta} \tag{3}$$

where σ_o is the strength of a hypothetical single phase alloy with a dislocation density so low that it has no influence on strength, α is a constant of order of magnitude unity, M is an orientation (Taylor) factor, G the shear modulus and b the Burgers vector. The composite flow strength is assumed to be a volumetric average of the flow stresses of the individual phases, i.e.,

$$\sigma_C = V_A \sigma_A + V_B \sigma_B \tag{4}$$

where the Vs are respective volume fractions and the respective σs are obtained from Eq. (3) via solution of Eq. (1) (which must be done numerically). To predict composite strengths requires, first of all, knowledge of the coefficients σ_o, C_1 and C_2 for each phase. These can be obtained by reasonably estimating α and M and determining the flow stress of the single phase materials as a function of deformation strain. A reasonable value of P_A (or P_B) is then assumed (e.g. $P_A = P_B = 0.5$) and then one strength datum for a two phase alloy (i.e. at one volume fraction and processing strain) allows the adjustable constant K to be ascertained. Hence our model is effectively a one parameter one which allows prediction of two phase strengths as a function of phase volume fraction and processing strain to be made solely on the basis of the flow behavior of the individual phases and one datum from a cold worked two phase solid. This is clearly one of the model's attractive features.

The degree to which our description is successful in predicting strengths is illustrated in Fig. 1 for HDISC fabricated in our, and other's, laboratories (5,10). As indicated by this figure, the model is clearly a robust and successful one in terms of its predictive capabilities, and in spite of deficiencies in microstructural detail.

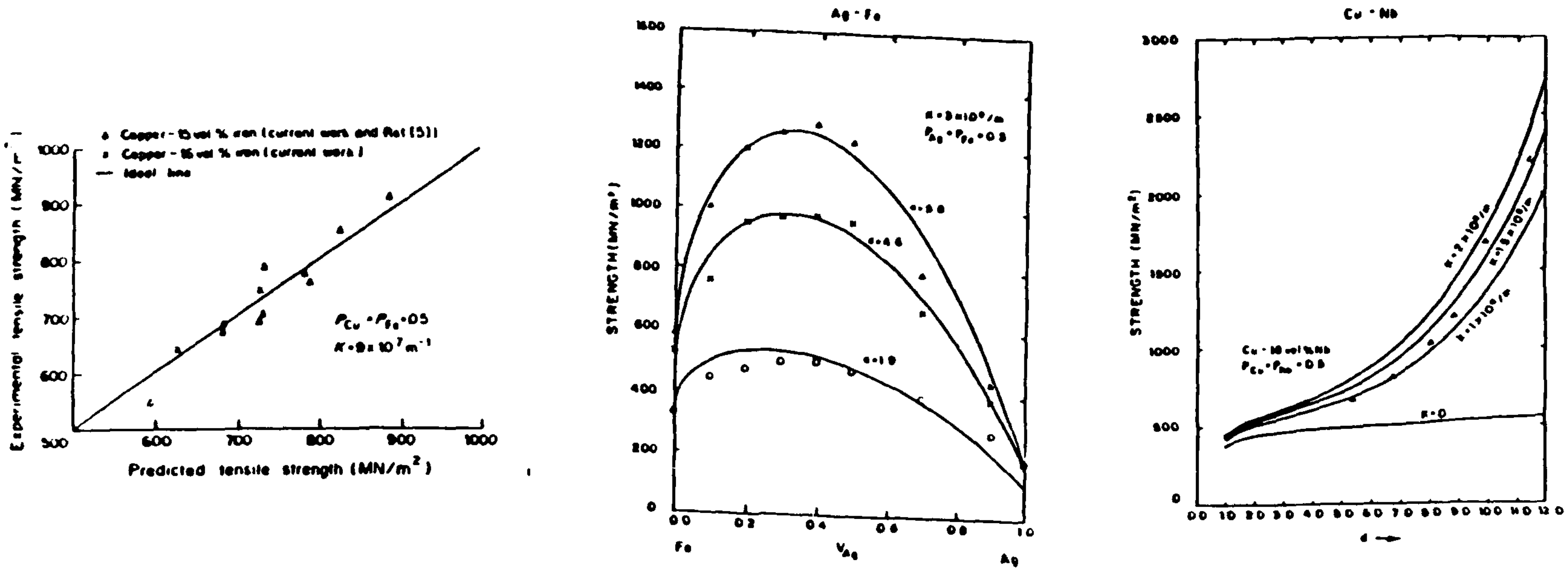
Several studies have directly (4) (via transmission electron microscopy) and indirectly (via resistivity) verified the development of high dislocation densities in HDISC. On the other hand, microstructural examination at the electron optical level of composites subjected to deformation strains on the order of ten or so have revealed regions of relatively low dislocation density in some HDISC (4). More recently, it has been reported that some HDISC are characterized by unremarkable dislocation densities even at relatively modest deformation strains (ϵ=5.0) (15,16). Studies along this line have been relatively sparse, and this may account for the apparent contradictory nature of these results. Thus while our model can be judged as an unqualified phenomenological success, more detailed and

comprehensive structural analysis is called for if we are to develop a fuller understanding of strengthening mechanisms in HDISC, thereby enhancing our capacity for further development of them.

III. Processing

Original studies on HDISC were restricted to metal combinations (e.g., Cu-Nb (4,6), Ag-Cu (3)) inherently ductile in single phase form. More recently it has been shown that some metal combinations (e.g., Ni-W (8,9), Cu-Cr (5,7), Cu-Mo (5)) containing a phase commonly thought not capable of such processing can also be extensively deformation processed. For example, arc melted Cu-Cr alloys (7) can be cold worked extensively - and Cr "whiskers" produced thereby - for alloys containing less than about 20% Cr by volume, even though polycrystalline chromium can not be cold worked at room temperature. However, Cr has the requisite number of independent slip systems to maintain strain compatibility across Cu-Cr interphase boundaries, and it is this feature that apparently allows Cu-Cr alloys to be cold worked. More specifically, in arc melted castings the Cr dendrites exist as isolated single crystals within the Cu matrix if the Cr volume fraction is less than about 20%; these are the alloys that can be extensively cold worked. However, when the Cr content exceeds 20%, Cr-Cr interparticle contacts are formed. These fracture during processing, and HDISC can not be made from castings containing these higher chromium percentages. Nickel-tungsten alloys, containing either isolated or continuous tungsten particles, can be deformation processed to true strains on the order of three to four (8,9). When the tungsten is continuous, the alloy must be first hot worked prior to cold deformation. During hot working, the W-W interparticle contacts fracture, but this is not concurrent with material failure. Instead, these cracks are "healed" through matrix flow, and the resulting single crystals of tungsten can then be cold worked rather like the isolated Cr dendrites in Cu-Cr alloys. Finally, Cu-Mo alloys behave somewhat similarly to Ni-W alloys during deformation processing. Polycrystalline molybdenum particles fracture intergranularly (7), but this does not lead to composite fracture if the molybdenum content is not too high. Following fragmentation, the resulting single crystals of Mo deform concurrently with the copper matrix. These studies show that there are some subtle mechanical and microstructural characteristics which determine the suitability of a given two phase combination to be made into a HDISC.

We have conducted composite processing studies having a technological focus during the course of this grant. We have studied, in particular, the fabricability of a number of Ni-W alloys of varying W volume fraction and produced by a variety of



(a) (b) (c)

Figure 1: The model for strengthening of two phase cold worked solids is useful for predicting strengths in (a) Fe-Cu, (b) Fe-Ag and (c) Nb-Cu. In (a) experimental and predicted strengths are compared for Fe-Cu alloys drawn to different strain levels and interphase spacings. In (b) we show how the model compares with strengths as they vary with relative phase volume fraction in Fe-Ag alloys drawn to several levels of strain, and in (c) the degree to which the model mimics the work hardening rate of a Nb-Cu alloy is illustrated. For the Nb-Cu alloy a value of $K=(1.0 \text{ to } 1.5)\times10^8/m$ gives the best fit to the data. (The curve marked K=O is the strength predicted on the basis of the work hardening behavior of the single phases.)

initial processing techniques. As indicated in Table 1, all of directional solidification, casting, liquid phase sintering and transient liquid phase sintering were employed to produce starting structures of Ni-W two phase alloys.

Directionally solidified eutectics of Ni-W, in which the W phase forms as blades or rods distributed within the solid solution matrix, can be cold drawn to deformation strains on the order of four. At this point, the composite has a tensile strength of 2400 MN/m (~350,000 psi) which precludes it from being drawn to greater strains by the techniques available to us. Reductions in area remain appreciable (ca. 20%) even at the highest strength levels, and this is evidence of the excellent combination of strength and toughness that can be developed in HDISC of these metal combinations.

During processing the transverse cross-sectional shape of the tungsten remains the same up to deformation strains on the order of three or so (that is, the tungsten remains as rods/blades, with transverse dimensions and mean interphase spacings decreasing commensurately with drawing strain). However at deformation strains approaching four, the tungsten begins to assume a chevron shaped cross-section (this is the shape characteristic of Fe and Cr in Cu-Fe and Cu-Cr HDISC). This shape change is indicative of microscopic plane strain deformation, but we do not know if there is a causal relationship between the increased work hardening observed at these strains and this change in tungsten morphology.

Directional solidification is a fairly cumbersome and, therefore, expensive initial processing scheme. As such it deters from the otherwise economically attractive scheme of making HDISC. For this reason we initiated work (9) using the other processing schemes listed in Table 1.

Transient liquid phase sintering of Ni-W powders can be effected with luck and/or precise control of thermal processing. When a mixture of nickel and tungsten powders is heated above nickel's melting point, the solid tungsten particles have a pronounced tendency to settle as a result of gravity effects. At the same time, tungsten is dissolving in the nickel liquid and a maximum melting point is found in this binary alloy (17) (T =1785 K, c=16 at.% W). The liquidus temperature of the nickel is increased with tungsten dissolution, and if saturation is achieved prior to appreciable tungsten settling, a fully dense and homogeneous nickel-tungsten solid results. We have (on occasion) been able to generate such structures, and these can be cold processed to deformation strains on the order of three or more.

When an ordinary casting of Ni-W is made by heating Ni and W powders in a mold, pronounced gravity settling takes place for reasons noted above. For a typical casting made in our

Table 1: Processing Schemes Applied to Ni-W 24.2 at% alloy

Initial Process	Subsequent Processing
Directional Solidification	Swaging/Drawing to True Strains of 4.0
Induction Melting and Solidification in Al_2O_3 molds in vacuum	Rolling to True Strain of ca 2.4 Following Homogenization Heat Treatment
Transient Liquid Phase Sintering of Powder Compacts at T=1723-1773K for 8h	Rolling to True Strain of ca 2.4 Following Homogenization Heat Treatment

laboratory this results in two types of structure. At the ingot bottom, where settling is found, the temperature remains above the eutectic and a typical liquid phase sintered structure is formed there (W volume fraction ca. 48%). This portion of the structure can be extensively cold rolled, provided it is preceded by an initial hot breakdown. During the latter, W-W particle contacts are fractured, but this does not lead to composite fracture. The resulting single crystals of tungsten can then be cold rolled in the composite. The upper portion of the ingot contains ca. 7 vol.% W. It is structurally characterized as a typical hypoeutectic alloy, and can be cold rolled with ease. The strengths of all of these more "conventionally" processed alloys are comparable to, or exceed, the strengths of the cold worked directionally solidified eutectic (9).

It is clear from these examples that there are subtle and incompletely understood microstructural features and material mechanical properties which determine the capability of a two phase solid to be extensively cold worked. Nonetheless these studies indicate there are reasonable expectations that the combinations of materials capable of such processing are not as restricted as was originally thought.

IV. Thermal Stability

One distinct advantage of metal matrix composites vis-a-vis polymeric ones is that metals can be used to much higher temperatures than organic ones. By virtue of the high strengths they exhibit, the high temperature stability of HDISC is suspect. In particular, it would be intuitively suspected that recovery/recrystallization would proceed more rapidly in them than in comparably deformed single phase materials. Moreover, shape stability of the filaments and plates is questionable as a result of surface energy driving forces, particularly at the fine dispersions accruing in HDISC at large deformation strains. In work conducted under this program we have conducted preliminary investigations regarding the resistance of Fe-Cu HDISC to recrystallization, and have conducted extensive investigations relative to the shape stability of these materials at elevated temperatures.

We have studied the recrystallization kinetics of heavily deformed Cu-15 vol. % Fe alloys by measuring the reduction in their room temperature tensile strengths following elevated temperature exposure of alloys subjected to several prior deformation strains (18,19). Three alloys were drawn from arc melted material to strains ranging from three to five; the corresponding interphase spacings varied from ca. 3.5 micrometers to 1.2 micrometers. An additional alloy was first cold worked and then heated to a temperature high enough (800°C) to completely recrystallize the iron and the copper and to effect spheriodization of the iron filaments. This alloy was then bundled and drawn to a strain of two at which point the

interphase spacing was ca. 0.25 micrometer. The recrystallization response of these materials is shown in Fig. 2 (it is of interest to note the cold worked strength of the "fine" composite; it exceeds that of iron cold worked to the same degree and indicates the substantial advantage of starting with fine microstructures insofar as generating high strengths in HDISC are concerned). We can assess the relative recrystallization rate of the composites by measuring the initial decrease in strength of them relative to the comparable changes in strength of single phase material. The ratio of these changes in strength are provided in Table 2. (a ratio less than one indicates the composite is softening at a less rapid rate than one would expect based on the softening behavior of the composite constituents). In none of the cases is the recrystallization rate of the composite greater than that of the single phase material and in some of the composites the recrystallization response is markedly reduced in comparison to the single phase recrystallization kinetics.

Our current belief is that the more heavily cold worked structures resist softening because continuous recrystallization supplants discontinuous recrystallization as deformation strain and microstructural fineness increase. However, this interpretation has not been validated by the necessary electron microscopic observations. Nonetheless the results of this work clearly indicate that HDISC can, in some cases at least, be used structurally to as high a temperature as their comparably deformed constituents.

We have also experimentally studied and theoretically modeled shape instabilities in Cu-Fe and Ni-W HDISC (19). Plate shaped reinforcements are common to both alloy systems, for even in cold drawn Fe-Cu composites the iron phase deforms in a microscopically plane strain manner as a result of crystallographic slip considerations (20-22). The result is a chevron shape fiber which can be approximated as a plate. Several different plate shape instabilities can form in the reaction path which eventuates in a spheroidal dispersion of particles. These are illustrated schematically in Fig. 3. As shown in Fig. 3a, a plate can directly evolve into a cylinder which will then decompose into a row of spheres via a Rayleigh instability. This direct cylinderization is favored for plates having low width to thickness ratios. In contrast, when this ratio is large a row of spheres will form along the plate edge (Fig. 3b). This edge spheroidization process comes about as a result of the formation of a ridge (pseudocylinder) along the plate edge, which then decomposes into a row of spheres in a manner somewhat analogous to a Rayleigh instability. Finally if internal plate boundaries are present as a result of recrystallization or deformation texture, their recession can lead to a boundary splitting kind of instability (Fig. 3c).

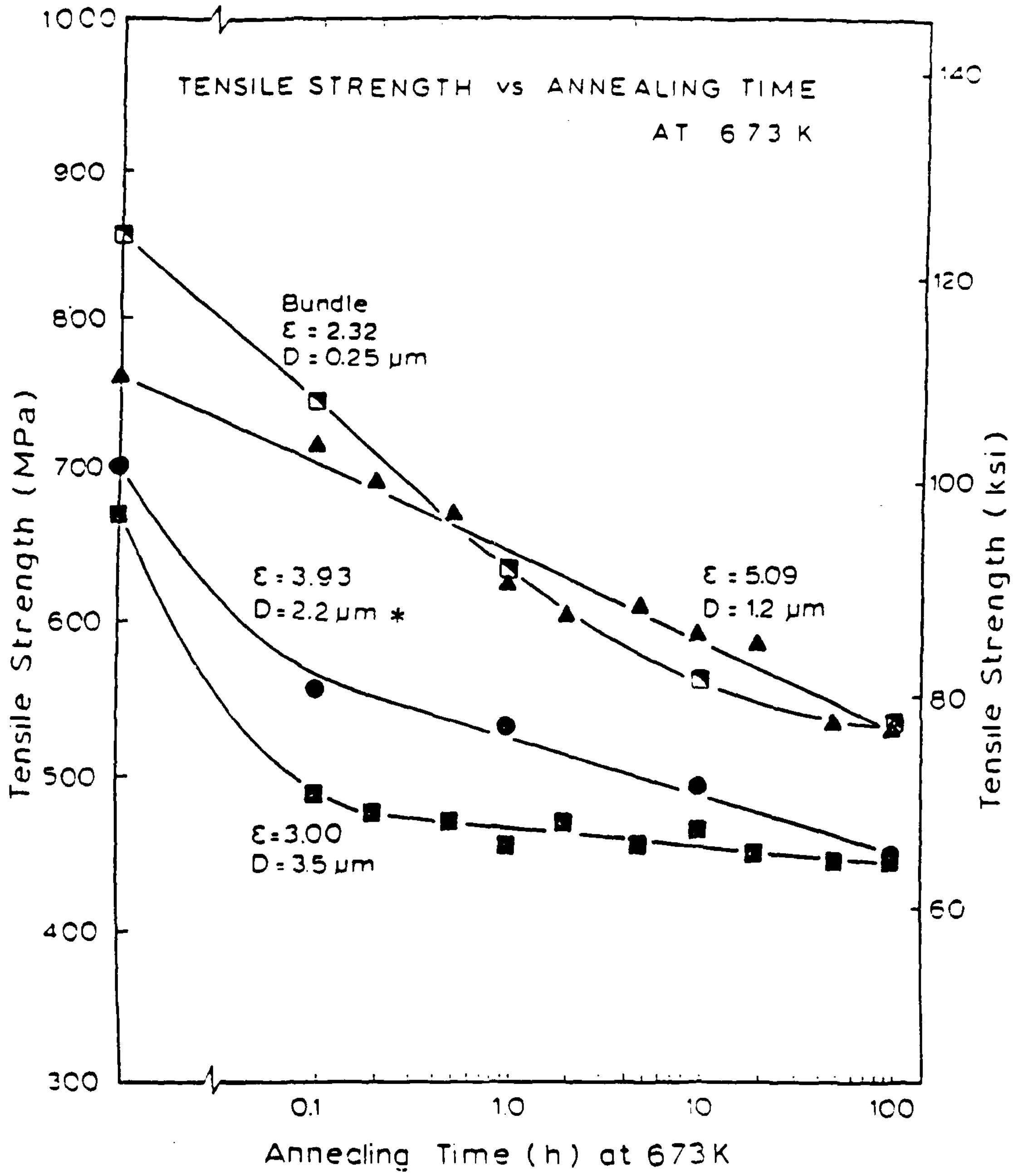


FIGURE 2. Combined results from single wire and bundle composites showing the influence of fiber spacing and cold work on annealing response. Data on the "Tensile Strength" axis represent cold worked strengths.

TABLE 2. Relative recrystallization rates, as measured by the ratio A/B, for Cu-13.4 wt.% (14.3 vol.%) Fe composites, where

$$A = \frac{(\sigma_{CW} - \sigma_{0.1})}{\sigma_{CW}} \quad \text{EXP-A}$$

and

$$B = \frac{(\sigma_{CW} - \sigma_{0.1})}{\sigma_{CW}} \quad \text{ROM-B}$$

σ_{CW} = Cold worked strength of composite/rule of mixtures respectively

$\sigma_{0.1}$ = Strength after 0.1 hr. at 693K for composite/rule of mixtures, respectively

ϵ	A/B			
	D = 3.5 μm	D = 2.2 μm	D = 1.2 μm	D = 0.25 μm
2.32				0.445**
3.00	0.944*			
3.93		0.620*		
5.09			0.170*	

* Single wire composites.

** Bundled composites.

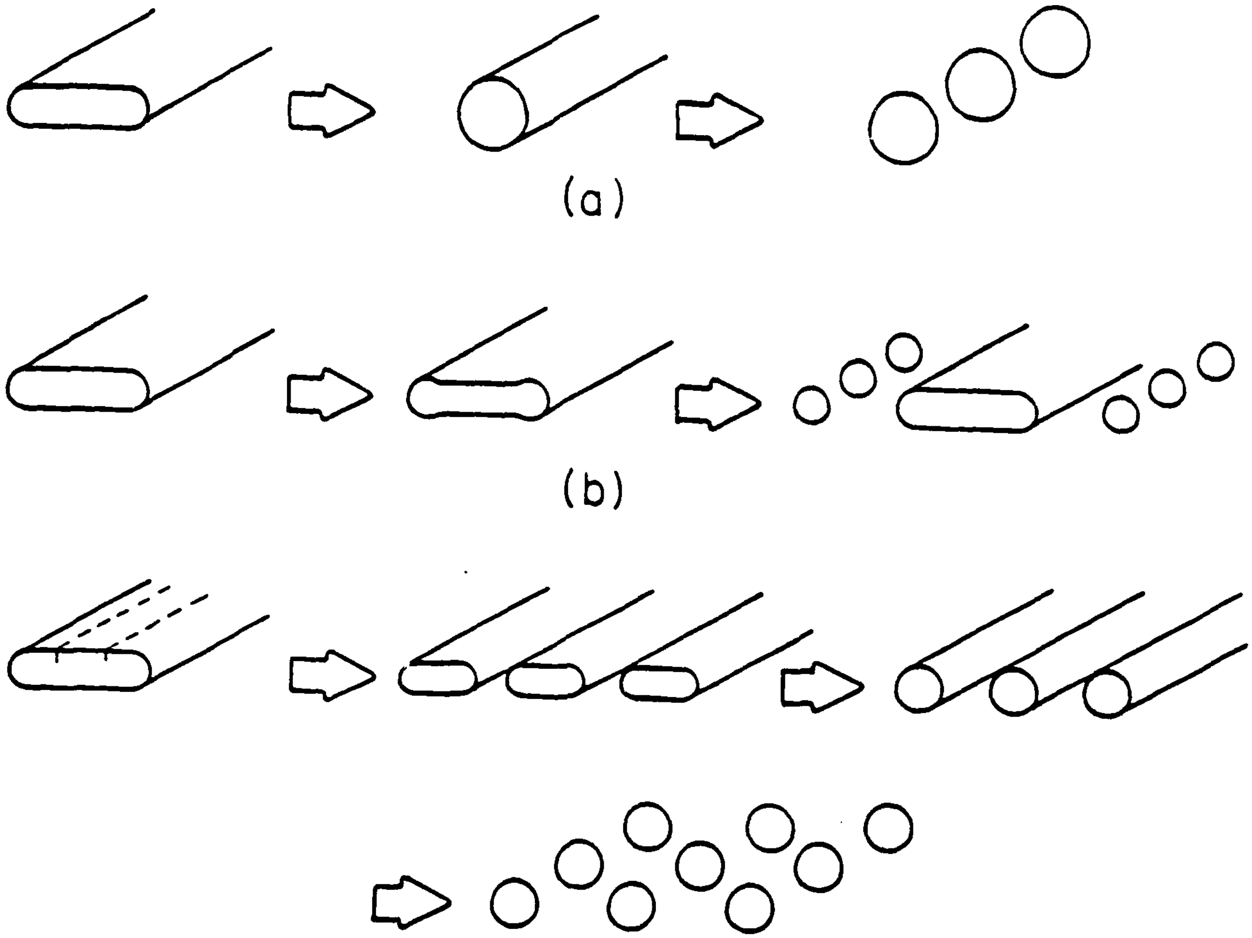
All three kinds of shape instabilities have been observed in the systems we have studied. The tungsten in rolled, liquid phase sintered Ni-W alloys first recrystallizes into a bamboo structure, following which the tungsten platelets split along these transverse boundaries (Fig. 4). Tungsten plates with low aspect ratios in hypoeutectic Ni-W alloys cylinderize directly and then undergo a Rayleigh instability (Fig. 5), whereas all of longitudinal boundary splitting, edge spheroidization and direct cylinderization are observed in Fe-Cu HDISC (Fig. 6).

We have modeled the various kinds of shape instabilities (19,23) in a way that allows predictive capabilities. In particular a shape instability diagram (Fig. 7) can be constructed in which the dominant mode of instability can be predicted in terms of the plate aspect ratio and the dihedral angle developed at internal boundaries. As indicated in Fig. 7, edge spheroidization is favored for plates with high aspect ratios and large dihedral angles, direct cylinderization is promoted when the aspect ratio is small and the dihedral angle is large and boundary splitting is the dominant instability mode when the dihedral angle is small. The predicted instability modes are in accord with those observed (24). Moreover, we have also modeled the transition from direct cylinderization to boundary splitting by a finite difference analysis (25), and the results of this more sophisticated approach are in accord with the analytical approach taken in development of the shape instability diagram.

V. Ancillary Studies

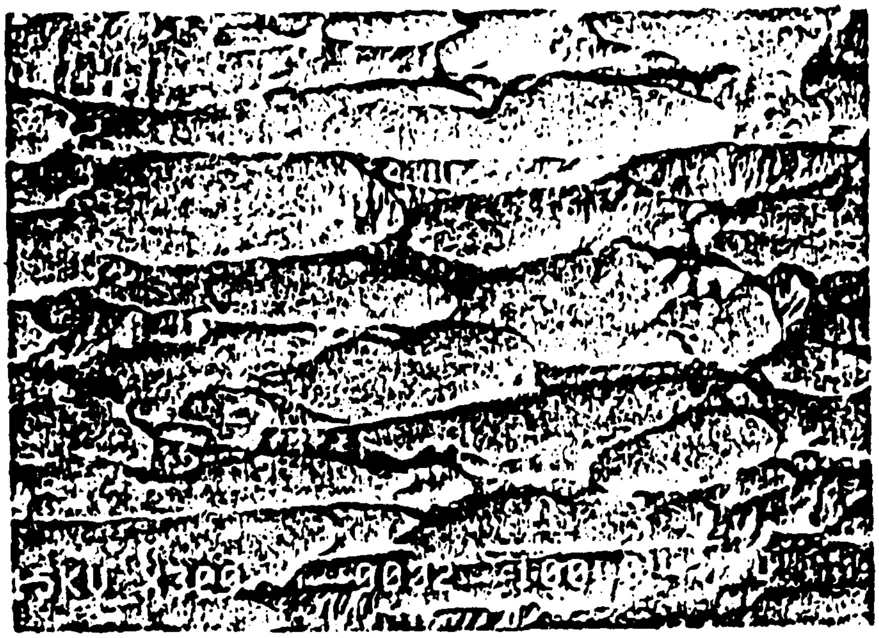
Other research has also been conducted under the aegis of this contract. In particular we were intrigued by the crystallographic requirements that must be met for two phase alloys to be rendered suitable for extensive deformation processing. Since most previous studies of HDISC had been carried out with cubic metals, we decided to investigate the suitability of fcc Al-hcp Zn for extensive deformation processing. However, this study was not completed in the way originally intended. Instead we found that the fabricability of this material was extraordinarily sensitive to initial processing conditions, and the presence of porosity and inclusions in the initial cast charge. As a result the ductility of Al-Zn alloys as a function of strain rate and temperature was investigated over a broad range of these variables and quantitatively linked to the initial defect density (26). Partial support for this portion of the program was provided by a grant for student support from the Alcoa Foundation.

We had also been intrigued by the possibility of using metallic glasses as a constituent for HDISC. Initial studies were carried out, though, on metal-metallic glass laminates (27). The metals used in the study were copper and brass; the ductility and strength of these constituents were varied by using them in



(c)

Figure 3. Various kinds of shape instabilities in plates. (a) A cylinder can form directly from a plate, and then decompose into a row of spheres. (b) During the process leading to cylinderization, a ridge first forms near the plate-end. A perturbation along it can lead to spheres forming along the plate edges. The process will repeat until such time as direct cylinderization takes precedence. (c) If a recrystallized high angle or recovered low angle boundary forms along the plate length, thermal grooving can cause the plate to split into N plates. These will then either directly fiberize or form spheres along their edge, depending on the new aspect ratio (w/Nt).



(a)

(b)

Figure 4. (a) An in plane view of a liquid phase sintered Ni-48 vol% W composite cold worked to a strain of 1.8 (b) the "plates" split along recrystallized boundaries after exposure to a temperature of 1673K for 5 hrs.

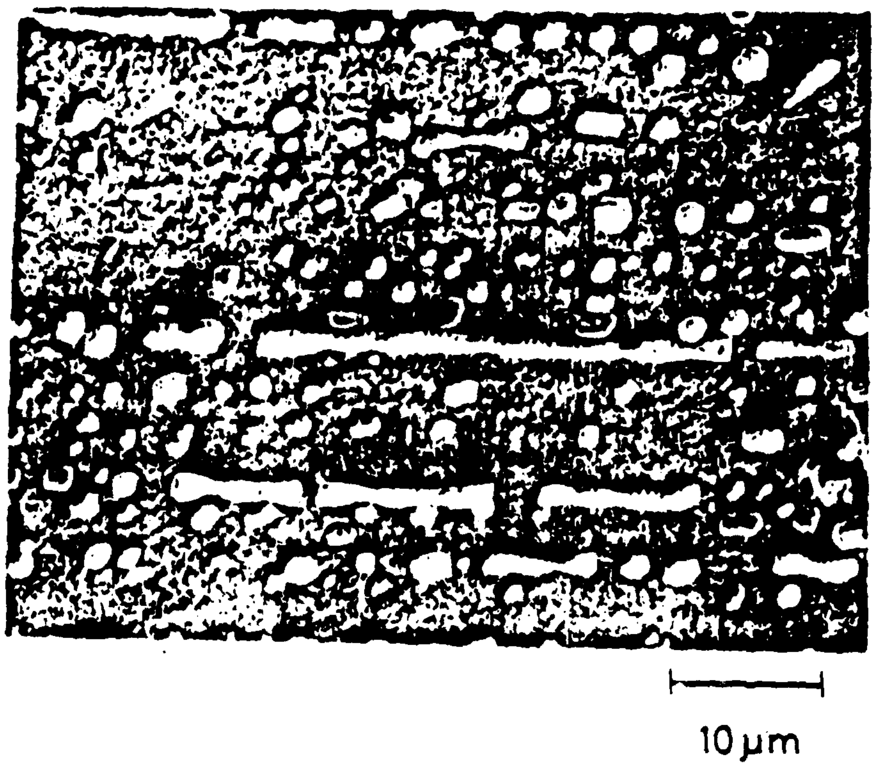


FIGURE 5. A deeply etched "in plane" section of a "hypoeutectic" Ni-W composite(ϵ = 1.7) following five hours at 1673 K. The remnants of cylinders evidence the cylinderization that precedes the breakup of cylinders into spheres.

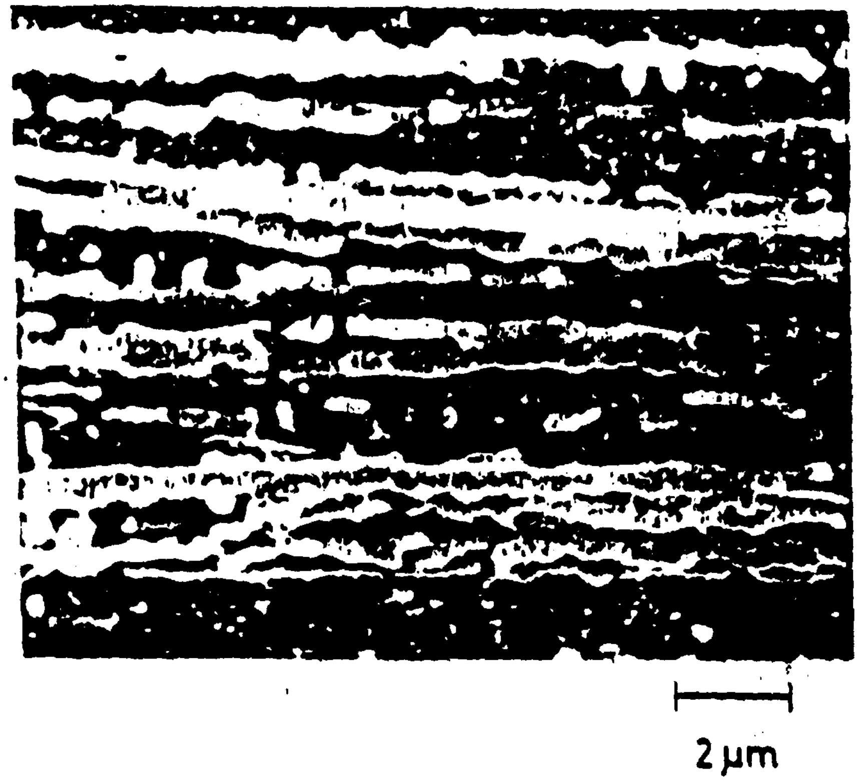


FIGURE 6. Evidence of the separation of small cylinders from the main ribbon via edge "spheroidization" mechanisms. Note the nodules along the same ribbon, to the left of the small cylinder, indicative of edge spheroidization. The central plate also indicates boundary splitting and there is also evidence of direct cylinderization of some of the plates. The micrograph is from a deeply etched longitudinal section of a Cu-13.4 wt.% (14.3 vol.%) Fe composite wire (ϵ = 5.09; D = 1.2 μm) following a two hour anneal at 1073 K.

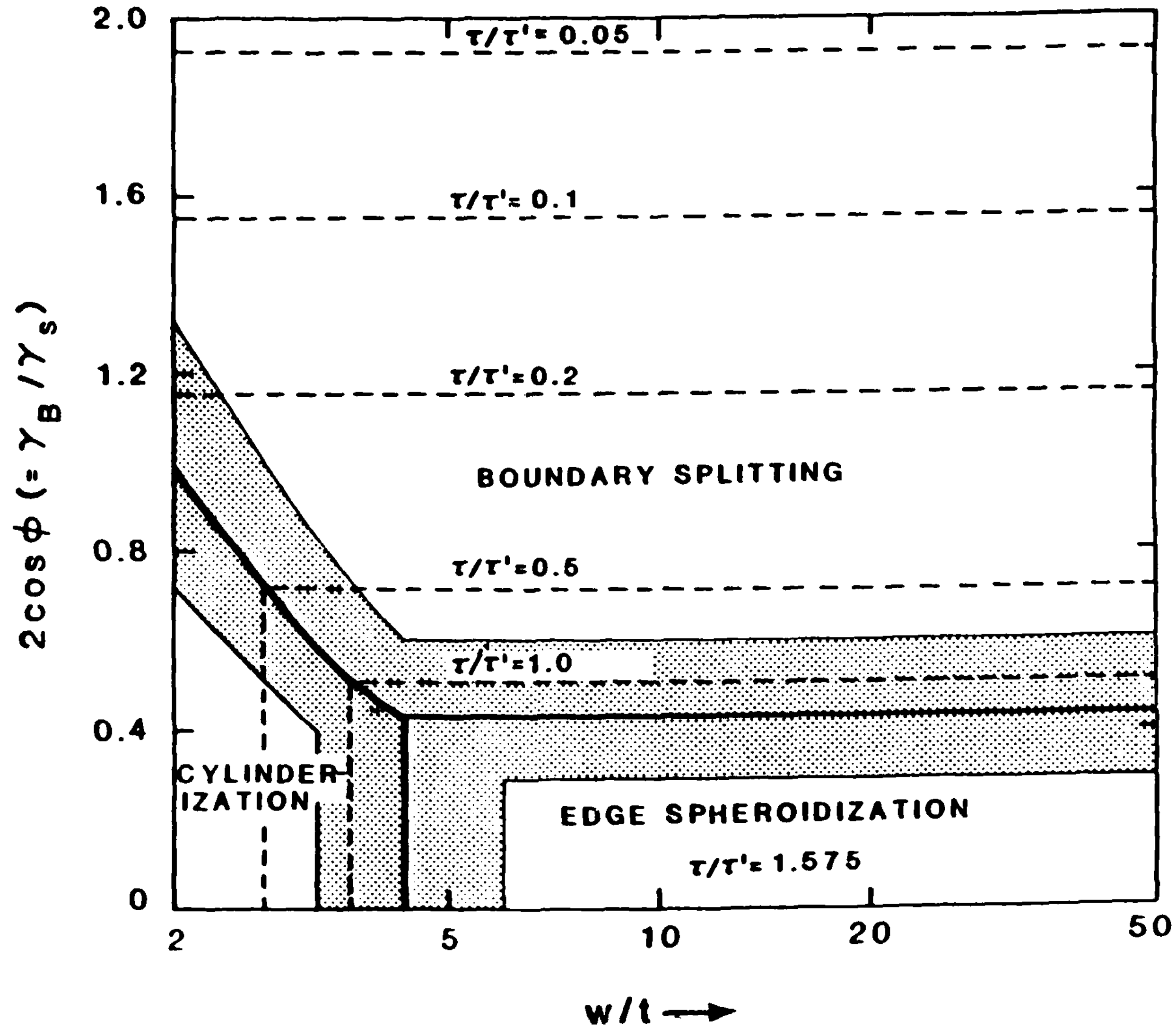


FIGURE 7. A plate instability diagram for volume diffusion controlled plate instabilities. The plate breakdown mechanism is indicated in the several regions of the diagram. Time for plate splitting (specified in terms of a base time) are superimposed on the diagram.

various degrees of cold work. We found that tensile ductility of these composites exceeded those of the metallic glass, sometimes significantly so. Moreover, multiple shear banding was observed in the metallic glass in the vicinity of the final fracture surface, indicating a beneficial effect of the metal in promoting fracture toughness in the glass. We were able to correlate the incremental increase in tensile ductility with the matrix ductility as indicated in Fig. 8. We are currently using fracture toughness and finite element modeling to determine the basic physics behind the toughening process.

VI. Conclusions

The following conclusions can be drawn from the results reported above.

1. A variety of two phase metals are suitable for manufacture into HDISC. Less malleable metals must be in a suitable morphological form for them to be ductile in two phase materials. In particular, their isolation as single crystals imbedded in a ductile matrix is most propitious for enhancing their malleability. In addition, the less ductile constituent must possess the requisite number of slip systems to provide for strain compatibility across interphase boundaries.

2. The strengths developed in HDISC can be predicted via a phenomenlogical equation which relates strength to inherent material properties, the phase volume fractions and a parameter which depends on the inherent flow behavior differences of the composite constituents. Even though deficient in microstructural details, the model is remarkably successful when used in a predictive manner.

3. HDISC are not particularly sensitive to elevated temperature softening. This is somewhat surprising in view of their high strengths and we believe this behavior is related to the fine dispersion characteristic of these kinds of materials. The dispersion serves to reduce the kinetics of discontinuous recrystallization.

4. Surface energy driving forces render HDISC shape unstable at elevated temperatures (temperatures greater than about two-thirds of the absolute melting temperature). The path to microstructural breakdown depends on plate geometry and the energy of any internal boundaries that might be present in the plate. Work in this program has delineated the kinds of instabilities that occur, and we have also been able to predict the path taken during breakdown in terms of parameters noted above.

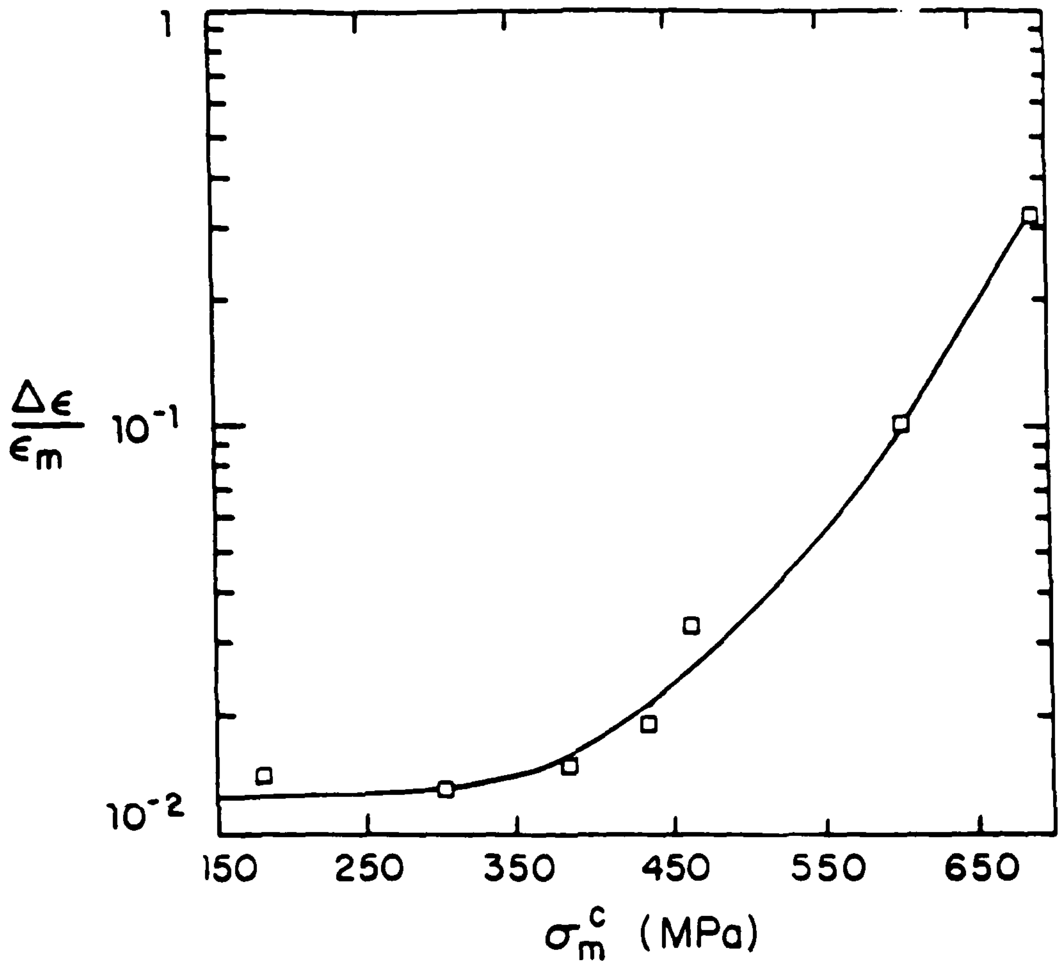


FIGURE 8. The increase in metal-metal glass composite ultimate strain relative to matrix tensile strain as it depends on matrix strength. Significant increases in tensile strain are associated with stronger metallic matrices

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24. J. C. Malzahn Kampe, T. H. Courtney and Y. Leng, manuscript in preparation, (1987).

25. J. K. Lee and T. H. Courtney, manuscript in preparation, (1987).

26. B. Harriprashad, T. H. Courtney and J. K. Lee, accepted for publication in Metall. Trans A.

27. Y. Leng and T. H. Courtney, manuscript in preparation, (1987).

Appendix A: Degrees granted under this program

David G. Kubisch, M. S., Michigan Technological University, May, 1985.

Yang Leng, M. S., Michigan Technological University, December, 1985.

Bhoopaul Harriprashad, M.S., Michigan Technological University, September, 1986. (Mr. Harriprashad had all of the requirements for the degree completed prior to his return to his native Guyana in October, 1986. He died tragically shortly after his return there, and before he orally defended his thesis. Mr. Harriprashad received partial support from the Alcoa Foundation.)

Jean C. Malzahn Kampe, Ph.D., Michigan Technological University, May, 1987.

David Maurice, M. S., University of Virginia (in progress).

Yang Leng, Ph.D., University of Virginia (in progress).

Appendix B: Publications resulting from this grant

1. P. D. Funkenbusch and T. H. Courtney, "On the Strengths of Heavily Cold Worked In Situ Composites", Acta Metall., 33, 913, (1985).

2. D. G. Kubisch and T. H. Courtney, "The Processing and Properties of Heavily Cold Worked Directionally Solidified Ni-W Eutectic Alloys", Metall. Trans. A, 17A, 1165, (1986).

3. J. C. Malzahn Kampe and T.H. Courtney, "Elevated Temperature Microstructural Stability of Heavily Cold Worked In Situ Composites", Scripta Metall., 20, 285, (1986).

4. P. D. Funkenbusch, J. K. Lee and T. H. Courtney, "Ductile Two Phase Alloys: Prediction of Strengthening at High Strains", Metall. Trans. A, 18A, 1249, (1987).

5. Y. Leng, T. H. Courtney and J. C. Malzahn Kampe, "High Strain Deformation Processing of Ni-W Alloys", Matls. Sc. and Eng., 94, 209, (1987).

6. B. Harriprashad, T. H. Courtney and J. K. Lee, "Porosity and Tensile Ductility in Al-Zn Alloys", accepted for publication in Metall. Trans. A.

7. J. C. Malzahn Kampe, T. H. Courtney and Y. Leng, "Shape Instabilities in Plate Like Structures: Part I: Experimental", in preparation.

8. T. H. Courtney and J. C. Malzahn Kampe, "Shape Instabilities in Plate Like Structures: Part II: Analysis", in preparation.

9. J. K. Lee and T. H. Courtney, "Two Dimensional Finite Difference Analyses of Shape Instabilities In Plates", in preparation.

10. Y. Leng and T. H. Courtney, "Tensile Behavior of Metal-Metallic Glass Laminates", in preparation.

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